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CHARACTERIZATION OF SUPERLATTICE DISLOCATIONS IN Cu_2NiZn BY TRANSMISSION ELECTRON MICROSCOPY.

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The aim of the present contribution is to extend our knowledge of order-disorder transformations in alloys by determining the dislocation spacing in a superlattice dislocation, which is related to ordering energy, stacking fault energy, etc.. If the slip systemis known, the separations between the partials of a superlattice dislocation can be calculated within the framework of anisotropic elasticity theory using values for the ordering energy and the stacking fault energy (1).

The Burgers vector of a dislocation can be determined using the criterion that a dislocation is invisible if $\vec{g}\cdot\vec{b} = 0$ and $\vec{g}\cdot\vec{b} \wedge \vec{u}$ is sufficiently small ($\lesssim 0.5$) (2), where $\vec{g}$ is the diffraction vector, $\vec{b}$ is the Burgers vector and $\vec{u}$ represents the unit vector along the superlattice dislocation line. In f.c.c. materials it is known that the Burgers vector is $\frac{1}{2}a_0$<110> and the slip plane is {111}. In Table 1 the <111> set of diffraction vectors are listed. Only three of the four different diffraction vectors are sufficient to determine the Burgers vector. It is essential to know whether a dislocation pair is a superlattice dislocation or a dipole. These can be distinguished by their dotted contrast or by tilting experiments (3). For proper observation in the case of a superlattice dislocation the foil can be tilted into the [110] orientation, so that the slip planes $(\bar{1}11)$ and $(1\bar{1}1)$ are perpendicular to the projection plane (110). Consequently, the images of the two unit dislocations belonging to the superlattice dislocation in such a plane coincide. Both (111) and $(11\bar{1})$ slip planes make an angle of 35° with the projection plane (110). A possible superlattice dislocation in those two planes will be seen separately as superpartials. The characterization of the slipplane can be done by tilting the foil towards one of the two possible slip planes and by observing the projected dislocation spacings. The results can be verified by stereo electron micrographs. The direction of the dislocation line can be calculated by determining the direction of the projected dislocation line [pqr] (see FIG.1). The plane perpendicular to the projection plane $(h_1\ k_1\ \ell_1)$ containing the projected dislocation line pqr is $(h_2\ k_2\ \ell_2)$. From FIG. 1 it is clear that $[h_2\ k_2\ \ell_2]$ is perpendicular to [pqr] and $[h_1\ k_1\ \ell_1]$. Hence we can write:

TABLE 1

$\vec{g}\cdot\vec{b}$ Values for <111> Reflections and $\frac{1}{2}\ a_0$<110> Burgers Vectors, 0: Invisible; $\pm$ 1: Visible.

$\vec{b}$ \ $\vec{g}$	[111]	$[\bar{1}11]$	$[1\bar{1}1]$	$[11\bar{1}]$
$\frac{1}{2}[110]$	+1	0	0	+1
$\frac{1}{2}[101]$	+1	0	+1	0
$\frac{1}{2}[011]$	+1	+1	0	0
$\frac{1}{2}[1\bar{1}0]$	0	−1	+1	0
$\frac{1}{2}[10\bar{1}]$	0	−1	0	+1
$\frac{1}{2}[01\bar{1}]$	0	0	−1	+1

$$[h_2\ k_2\ \ell_2] = [pqr] \wedge [h_1\ k_1\ \ell_1]. \qquad 1$$

The direction of the dislocation line is the secant of the plane $(h_2\ k_2\ \ell_2)$ and the slip plane $(h_3\ k_3\ \ell_3)$, resulting in:

$$[uvw] = [h_2\ k_2\ \ell_2] \wedge [h_3\ k_3\ \ell_3]. \qquad 2$$

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The dislocation spacing, d , of a superlattice dislocation can be obtained from the spacing measured from the projected two-dimensional image, d_I, using the following equation (4):

$$(d/d_I)^2 = \cos^2\alpha/\cos^2\Theta, \qquad 3$$

where:

- d = the real dislocation spacing,
- d_I = the dislocation spacing in the image plane (≡ projection plane),
- α = the angle between the projected and the real dislocation line direction,
- Θ = the angle between the slip plane and the image plane.

Strips of Cu_2NiZn are rolled to a thickness of about 35 μm, 4 mm wide and 12 cm long. These strips are encapsulated under vacuum in a narrow tube of fused quartz and annealed at 850°C for 4 days. Subsequently, these strips are annealed for 1000 h. at a temperature of 350°C in a salt bath and quenched in water. So the alloy is fully ordered and has an average domain size of 30 nm. After these thermal treatments the composition of the strip is 51 at.% Cu, 24 at.% Ni and 25 at.% Zn. The dislocations are introduced by a 5% elongation of the strip. A disc-type specimen is electro-chemically thinned by a jet method at room temperature in a 33 vol.% solution of concentrated orthophosphoric acid in water. Subsequently, electro-chemically polishing is carried out at -70°C in a 33 vol.% solution of concentrated nitric acid in methanol. The electron microscope is a Philips EM 300 and is operated at 100kV.

Almost all grains have a [110] orientation near the foil normal, due to rolling texture. Therefore, not all diffraction vectors of Table 1 can be obtained. Still, the Burgers vector can be determined by using also diffraction vectors of type <200> and <220>. FIG. 2 shows a number of weak beam images of dislocation pairs projected on the (110) plane using the diffraction vector $\vec{g} = [2\bar{2}0]$. Dislocation pair A is a superlattice dislocation with Burgers vector $\vec{b} = \frac{1}{2} a_0 [1\bar{1}0]$ and slip plane $(11\bar{1})$. From FIG. 2 we see that the projected dislocation line direction is $[1\bar{1}0]$. The projection plane $(h_1\ k_1\ \ell_1)$ is equal to (110) and the slip plane $(h_3\ k_3\ \ell_3)$ is $(11\bar{1})$. Using equations 1 and 2 : $(h_2\ k_2\ \ell_2)$ is equal to (001) and [u v w] is $[\bar{1}\bar{1}0]$. In conclusion we state that the dislocation has pure screw character. The projected dislocation spacing in FIG. 2, d_I, is 11.6 ± 1.1 nm. From equation 3 it follows: d=14.2 ± 1.4 nm. However, the image peaks do not coincide with dislocation core positions. This can be corrected using (5):

$$d = (d_c^2 + 4/C^2)^{\frac{1}{2}}, \qquad 4$$

where:

- d_c = core spacing of a superlattice dislocation,
- $C = -S_g/(\vec{g}\cdot\vec{b}/2\pi)$ for a screw dislocation,
- S_g = deviation parameter from exact Bragg position.

Applying equation 4 to the superlattice dislocation A of FIG. 2, the core spacing, d_c, is found to be equal to 12.9 ± 1.4 nm. The same dislocation is imaged on the $(11\bar{1})$- and the (010)-projection plane. A similar calculation gives:

$$d_c\ (11\bar{1}) = 12.0 \pm 1.1 \text{ nm},$$

$$d_c\ (010) = 10.7 \pm 1.5 \text{ nm}.$$

So the core spacing of a screw superlattice dislocation in fully ordered Cu_2NiZn is found to be 11.9 ± 1.1 nm. Partials of the two unit dislocations could not be detected with a magnification up to 93.000 X. In a theoretical model of a symmetrical screw superlattice dislocation De Hosson et al. (1) found a separation between the partial dislocations of 1.9 nm. The half-width of the peak, Δx, of a screw dislocation profile can be calculated from (6):

$$\Delta x = \frac{2.8}{\pi^2|S_g|} \approx \frac{0.28}{|S_g|} \quad . \qquad 5$$

Substituting $S_g = 0.1\ nm^{-1}$ in equation 5 , we obtain $\Delta x \approx 2.8$ nm. Hence, it is not possible to observe the partial dislocations using the above-mentioned value of S_g.

Kurdyumov and Chupyatova (7) have found superlattice dislocations in Cu_2NiZn with a dislocation spacing of 15.0 - 17.0 nm (uncorrected for the image displacement of the core). It is not clear whether these values are for edge or screw dislocations. At any rate, the value found in the present work is smaller.

A more extended research on the system Cu-Ni-Zn has to be undertaken. For instance the dislocation spacing has to be determined as function of the degree of ordering for screw, edge and mixed superlattice dislocations and more generally in relation to mechanical properties.

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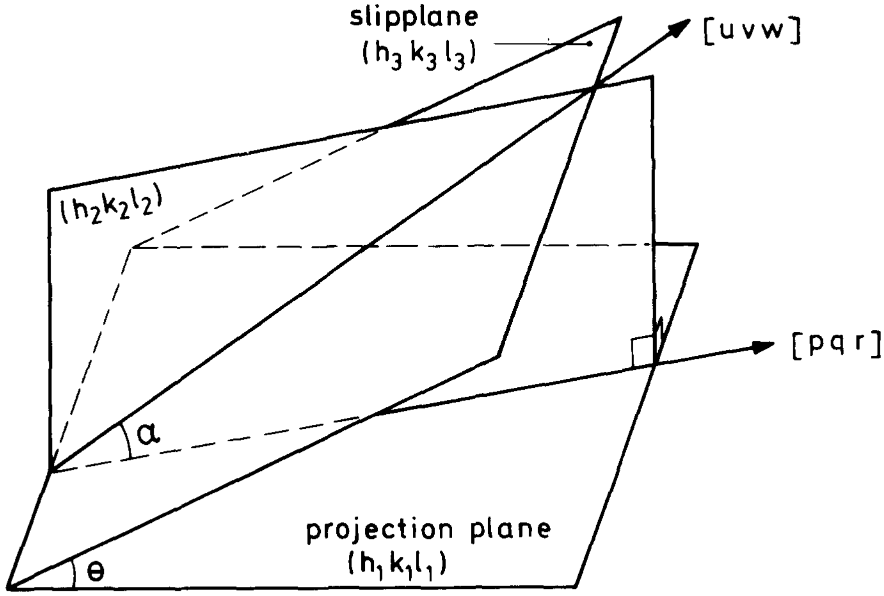

FIG. 1

Schematic representation of the projection of a dislocation line.

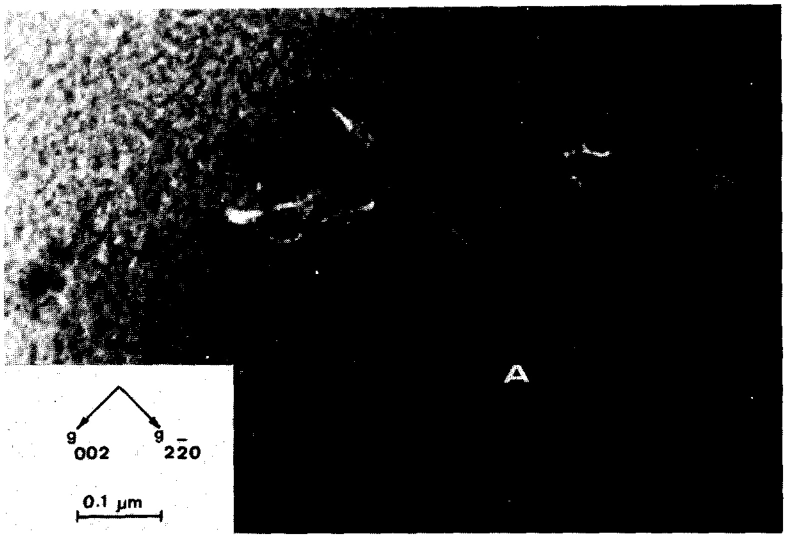

FIG. 2

Weak beam dark-field image of dislocation pairs; image plane = (110); $\vec{g} = [2\bar{2}0]$; $\vec{b} = \frac{1}{2}\, a_0[1\bar{1}0]$; $S_{\bar{2}20} = 0.1\ nm^{-1}$; dislocation pair A is a superlattice dislocation of pure screw character with a projected spacing of 11.6 nm.